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DEFORMATION BEHAVIOR OF AN AGE-HARDENABLE BETA + ALPHA-TWO TITANIUM ALLOY

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ABSTRACT

The beta Ti alloy Ti-23Nb-11Al (at. pct.) differs from other beta Ti alloys in that it is age-hardenable due to the formation of lath-like, alpha-two precipitates based on Ti_3Al. This study reports the room temperature deformation and slip behavior of the Ti-23Nb-11Al alloy as a function of heat treatment. The results indicate that the formation of the alpha-two precipitates results in not only a large increase in yield stress but also can induce a change in slip mode from inhomogeneous to uniform slip.

INTRODUCTION

Existing age-hardenable beta Ti alloys depend on the formation of hcp alpha-phase precipitates during the aging of solution-treated material in the 450°-500°C range; see ref. 1, for example. While a range of phase transformations are possible in metastable beta Ti alloys [2], most result in limited ductility or embrittlement, usually due to extremely non-uniform slip. An exception is the behavior of a Ti-7Mo-16Al (at. pct.) alloy aged at ≅ 900°C in which case coarse alpha-two particles in a disordered beta matrix cause a transition to wavy, uniform slip accompanied by an increase in ductility [3]. However, aging at lower temperatures resulted in multiphase microstructures in which ordered B2 and, in some cases, alpha-phase particles were present and embrittlement occurred due to inhomogeneous, planar slip [3,4].

In the present study, we report results on the deformation behavior of a new age-hardenable beta Ti alloy Ti-23Nb-11Al (at.pct.), hereafter referred to as Ti-23-11. This alloy remains disordered upon quenching and readily forms alpha-two particles upon aging at 575°C [5]. The result is pronounced age hardening from temperatures as low as 375°C to as high as 675°C, where an increase of hardness of ≈120kg/mm^2 still occurs [5]. The present study reports the room temperature stress-strain as well as slip behavior of Ti-23-11 compression specimens deformed after the following heat treatments: (1) solution-treated (ST) obtained by quenching from 1000°C, (2) underaged in which ST material was aged at 375°C for 6 hrs., (2) peak-aged which consisted of ST+ 575°C/6 hrs, and (4) an overaged condition resulting from ST + 675°C/6 hrs.

EXPERIMENTAL

The Ti-23-11 alloy had a composition of Ti-22.8Nb-11.1 Al in at. pct. or Ti-38Nb-5Al in wt. pct. The material contained about 1580 wt. ppm oxygen, 160 wt. ppm nitrogen, and 11 wt. ppm hydrogen. Test specimens were machined from bar stock which was extruded at 1038°C using a 22:1 extrusion ratio. All heat treatments were performed by encapsulating Ta-wrapped specimens in quartz under a partial pressure of high purity argon. The compression specimens were electropolished at -40°C in a methanol, ethylene glycol, perchloric mixture [6] prior to aging.

Mechanical tests were performed at room temperature in compression on cylindrical specimens 6.35 mm in diameter and length at an engineering strain rate of $2x10^{-4}$/s. Molydisilicide lubricant was used on the end faces to reduce friction; recent results indicate that this technique effectively produces accurate stress-strain data in compression to strains $\geq$0.5 [7].

RESULTS AND DISCUSSION

1. Microstructure

The general features of the microstructures of the solution-treated (ST) condition as well as after aging at 575°C for 6 hours are described elsewhere [5]. The observations include the following: (1) upon quenching from 1000°C, the Ti-23-11 retains a disordered bcc lattice; neither athermal omega or B2 phase were detected, (2) upon aging at 575°C, lath-like alpha-two precipitates form with a zig-zag morphology; the precipitates are crystollographically related to the matrix with the Burger's orientation relationship and have an average length of ≅ 170 nm.

Figure 1 illustrates the contrast in precipitate size after aging at 375°, 575°, and 675°C; the differences in magnifications should be noted. After aging at 375°C for 100 hrs, a uniform distribution of very fine, 5-10 nm long precipitates is evident. Bright field imaging of the alloy in this condition suggests that the precipitates appear to have a distinct alignment with each other, probably consistent with the zig-zag morphology seen at 575°C. Although the selected area diffraction patterns (SADP) after aging at 375°C precluded proper indexing, those data are consistent with SADP's obtained after aging at 575°C, where the alpha-two structure is confirmed.

After aging at 675°C for 100 hrs, the lath-like precipitates have coarsened to a maximum dimension of about 500 nm with a minor dimension in the 50-100 nm range; see Figure 1c. While they still retain a zig-zag morphology, the interparticle spacings have increased significantly, suggesting a considerably lower strength. On the other hand, growth of the precipitates from approx. 170 nm after 575°C/6 hrs aging to roughly 500 nm after 675°C/100 hrs. suggests coarsening is relatively slow in this system.

2. Precipitation Hardening: Yield Strength

The microstructures shown in Figure 1 suggest that the Ti-23-11 alloy should be age hardenable. As shown in Figure 2, there is a large precipitation hardening response upon aging in the temperature range of 375°C-675°C. It should be noted that, given the six-hour heat treatment, the alloy is underaged at 375°C and overaged at 675°C. The 575°C/6 hr. represents peak hardness, and the yield stress value of ~1270 MPa is about 700 MPa greater than that in the solution-treated condition.

While we were not able to determine the volume fractions and interparticle spacings of alpha-two particles, the large age-hardening response shown in Figure 2 indicates the effectiveness of alpha-two precipitates in hindering dislocation motion. In the case of "conventional" beta Ti alloys, age hardened by alpha-phase precipitates, the alpha phase is known to be initially softer than the matrix; upon deformation, small alpha-phase particles work harden very rapidly [3,8]. As a result, particle hardening is achieved in large part by relying on the strain hardening of individual alpha-phase particles. In contrast, particle hardening based on alpha-two precipitates introduces an additional strengthening component due the long range order present in the alpha-two particles. Given the Burgers orientation relationships between the beta matrix and the alpha-two precipitates [5] and the presence of Nb in the precipitates,* we expect that, at small particle sizes, the alpha-two particles will be sheared by slip on the basal planes. Because of the ordered structure, basal slip within Ti_3Al-Nb alpha-two phase should occur by the motion of coupled pairs of $^a/_6$ $<11\bar{2}0>$ dislocations [9,10]. Thus, under conditions where shearing of precipitates is likely [underaged as well as peak-aged conditions], the particle shear process must either preserve long range order by superlattice dislocation motion or create an anti-phase boundary (which would have very high energy [10]) within the alpha-two precipitates. In short, strengthening due to ordered precipitates occurs. Alpha-two precipitates should thus be more effective in strengthening beta Ti alloys than disordered alpha-phase particles.

3. The Effect of Age-Hardening on Slip Mode

The fine scale of the precipitates, especially after aging at 375°C ($\lesssim$ 10 nm) suggests that upon yielding, dislocations should shear the precipitate particles, despite their ordered structure. Inhomogeneous (usually) planar slip is characteristic of precipitation-hardened beta Ti alloys when particle shearing occurs (see ref 3, for example). As Figure 3 shows, the Ti-23-11 alloy is no exception. Coarse, inhomogeneous slip is evident in the solution-treated condition as well as after aging at either 375°C or 575°C.

The slip lines in Fig. 3 tend to be planar although some cross slip is evident; furthermore, no significant change in slip line spacing was defected between the ST, ST + 375°C, and ST + 575°C conditions. Bcc alloys with large levels of solid solution hardening elements, such as in the solution-treated Ti-23-11, are known to exhibit coarse, planar slip [11]. Furthermore, the addition shearable precipitates is also known to favor inhomogeneous deformation in the form of continued

* The presence of Nb in Ti_3Al is known to promote basal slip at room temperature in Ti_3Al-base alloys [9].

slip on the same plane, given the slip-induced reduced-cross section area of the precipitate which intersects a particular slip plane. Thus, like most metastable beta Ti alloys in a hardened condition [2,12,13], Ti-23-11 is susceptible to coarse, inhomogeneous slip so long as particle shearing occurs. As expected, this is the case in the Ti-23-11 alloy from underaged to peak-age conditions; see Figure 4a to 4c.

In contrast to conventional beta Ti alloys, the Ti-23-11 alloy can be heat treated such that slip is so uniform and on such a fine scale as to be unresolvable even at 5 k in a SEM. This is shown in Figure 3c. Given the previous discussion regarding the difficulty of shearing the ordered alpha-two particles, it is reasonable to expect that, given the particle growth that occurs in the overaged condition, such as at 675°C, the particle size should increase to a level where the dislocations bypass the particles by looping. In the present study, TEM of the 675°C/6 hr condition was not performed. However, after 100 hrs at 675°C, the alpha-two particles grew to ≅ 50-100 nm compared to ≅ 30 nm after 6 hrs at 575°C (the dimensions quoted are the minor dimensions along which particle shearing is likely). As is well known, changing the dislocation glide mechanism from shear to by-pass of the precipitate particles can cause a dramatic change in slip distribution at low temperatures. This was demonstrated in Ti-Al alloys [14], in which homogeneously distributed fine slip occurs when large Ti_3Al precipitates (≅ 70 nm diam.) are looped but inhomogeneous, coarse slip results when smaller particles, 25-50 nm diameter, are sheared. In the present case, we conclude looping occurs when the alpha-two particles grow to 50-100 nm but that particle shearing persists for the ≅ 30 nm particles. Thus, the observations for the two alloy systems, both hardened by alpha-two particles but with much different matrixes, are remarkably similar.

4. Strain-Hardening Behavior

A consequence of the above dislocation-particle interactions is that, under those conditions where dislocation bypass of the particles occurs, a rapid rate of initial work hardening is expected from the accumulation of dislocations at the particles as geometrically necessary dislocations [15].

The extra work hardening $\Delta\sigma$ due to the accumulation of dislocations at particles may be expressed as [15]

$$\Delta\sigma = M^{3/2}\, C\, G \left(\frac{df}{d}\right)^{1/2} \sqrt{\varepsilon} \qquad (1)$$

where M is the Taylor factor which for bcc Ti is about 3, G is the shear modulus of beta Ti, b its Burger's vector, f is the volume fraction of precipitate particles, d is the average particle diameter, and ε is the strain. The term C is a constant whose predicted value is 0.24 [15].

Figure 4 shows that all of the alloys exhibit stress-strain behavior which is characterized by a short period of rapid initial hardening followed by extended linear work hardening in all conditions except the peak-aged (575°C/6hr). The transition to linear "Stage IV" work hardening occurs at very small strains ($\varepsilon \cong 0.03$) in the solution-treated (S/T) as well as the underaged (375°C/6 hr) conditions but is delayed to $\varepsilon \cong 0.08$ in the overaged case (675°C/6hr).

If, as we expect, particle looping occurs in the overaged condition, Eq. 1 indicates that parabolic work hardening should occur such that $\Delta\sigma \alpha \sqrt{\varepsilon}$. Fig. 5 shows this in fact occurs over the strain range $0.005 \leq \varepsilon$ 0.03. A value of the constant C may be estimated assuming a volume fraction of precipitate at 675°C after 6 hrs, f $\approx$0.1 [16]* and a particle diameter, $\approx$75 nm, intersecting the slip plane. In the term $\Delta\sigma$, we also take into account the intrinsic work hardening of the matrix. Such a calculation yields $C \simeq 0.31$, which is consistent with other observations for particle-hardened bcc alloys (C = 0.30 to 0.52 [17-19]) as well as being very close to the predicted value of 0.24 [15]. Thus we conclude that in the overaged condition, the rapid initial work hardening behavior is quite consistent with expectations based on dislocation bypass and the accumulation of geometrically necessary dislocations.

Linear strain hardening, usually denoted Stage IV, has been observed in both fcc and bcc metals [20-22]. However, significant differences exist between those observations and the present

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* This value of f was determined for an overaged condition of 850°C/180 hrs. Given the greater driving force for precipitation but decreased kinetics at 675°C, it should also be a reasonable estimate for the 675°C/6 hr condition.

data in Fig. 4. Typically, Stage IV occurs at much larger strains ($\varepsilon \gtrsim 0.5$) than those ($\varepsilon \simeq 0.1$) observed here. This may be a consequence of the high flow stress of the Ti-23-11 alloy and the suggestion that the onset of Stage IV occurs at smaller strains with increasing flow stresses [23]. The same analysis also predicts the magnitude of the linear work hardening $(d\sigma/d\varepsilon)_{IV}$ to increase with increasing flow stress [23]. However, Fig. 4 indicates $(d\sigma/d\varepsilon)_{IV}$ decreases with increasing flow stress; this has also been observed in Al alloys. In that case, the amount of solute in solution was believed to hinder the dislocation recovery process, causing an increase in $(d\sigma/d\varepsilon)_{IV}$ under conditions where substitutional solute content is high, such as in the ST condition [24]. This may occur in the present case as well, especially in view of the inhomogeneous slip in Ti-23-11 in all but the overaged condition. Finally, the typically observed values of Stage IV work hardening, $(\delta\sigma/d\varepsilon)_{IV} \simeq 1\text{-}2\times10^{-3}G$ [20,25]. The current data in Fig. 4 indicate $(d\sigma/d\varepsilon)_{IV} = 5$ to $25\times10^{-3}G$ or much higher than is commonly observed for Stage IV linear hardening. Some of this may be due to the inhomogeneous nature of slip and its influence of the recovery processes which should decrease the linear hardening rate. However, it is worth noting that both the peak-aged and overaged conditions exhibit similar $(d\tau/d\gamma)_{IV}$ values but much different slip characteristics. Thus, the slip mode cannot be a controlling factor in determining the linear hardening behavior of this beta Ti alloy. We conclude that, while this beta Ti alloy exhibit linear strain hardening at relatively small strains, it is not clear whether or not it is Stage IV work hardening, as often observed in fcc metals.

SUMMARY

A strong age-hardening response has been identified in the beta Ti alloy Ti-23-11. The hardening, which depends on solution-treating and aging heat treatments, is caused by the formation of fine-scale, lath-like alpha-two precipitates based on Ti_3Al. We propose that the ordered structure of the precipitate particles makes them more effective than disordered alpha phase precipitates in strengthening the matrix. Furthermore, it is possible to grow the alpha-two precipitate to a size (≈75 nm, minor dimension) wherein dislocation looping occurs. This results in a transition in slip mode from inhomogeneous to extremely uniform slip accompanied by rapid initial work hardening due to the accumulation of geometrically necessary dislocation. Linear strain hardening is observed at larger strains ($\varepsilon > 0.03$ to 0.10) but is not well understood.

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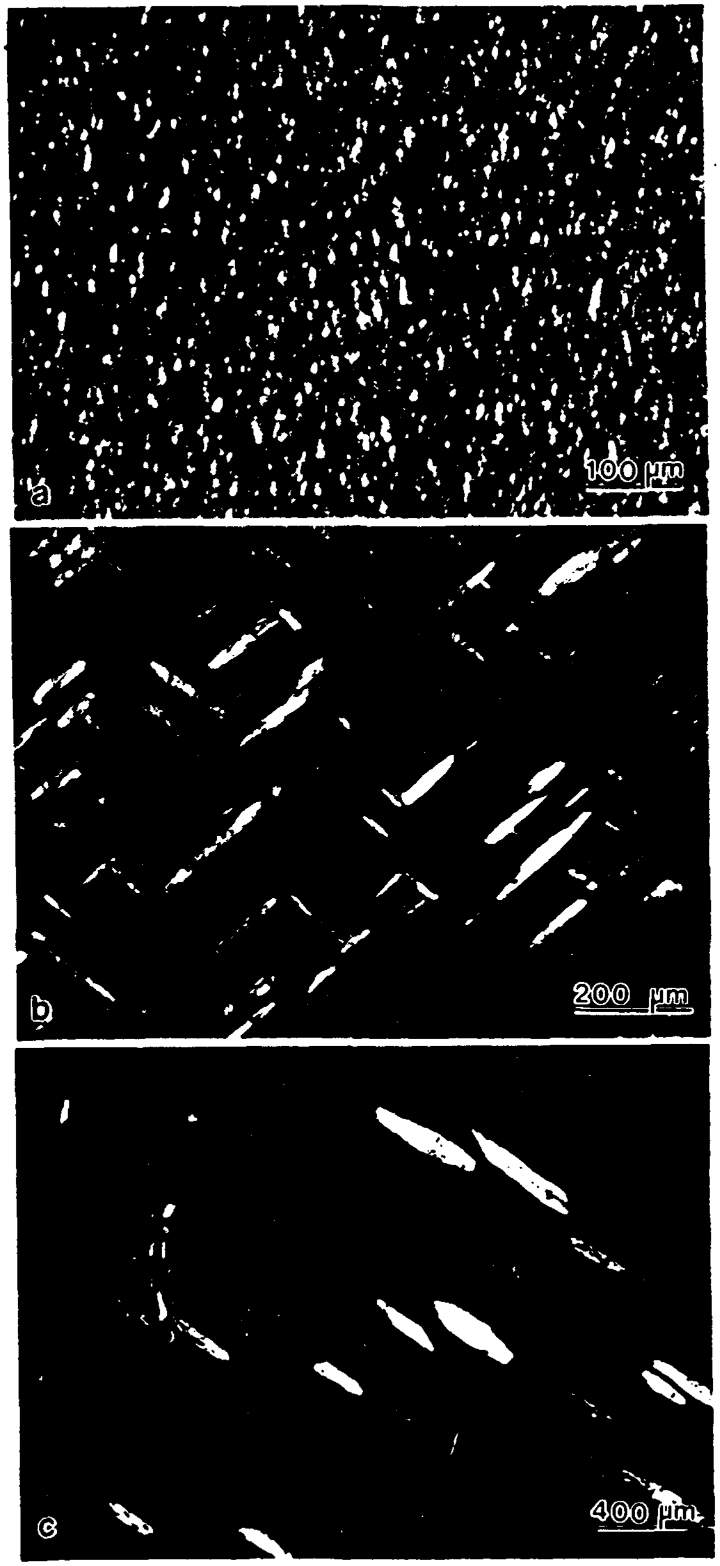

Figure 1. Centered dark-field image of precipitates in the Ti-23 Nb-11Al alloy quenched from 1000°C and aged at (a) 375°C for 100 hrs, (b) 575°C for 6 hrs, and (c) 675°C for 100 hrs. At 575° and 675°C, the precipitates have been identified as the alpha-two phase.

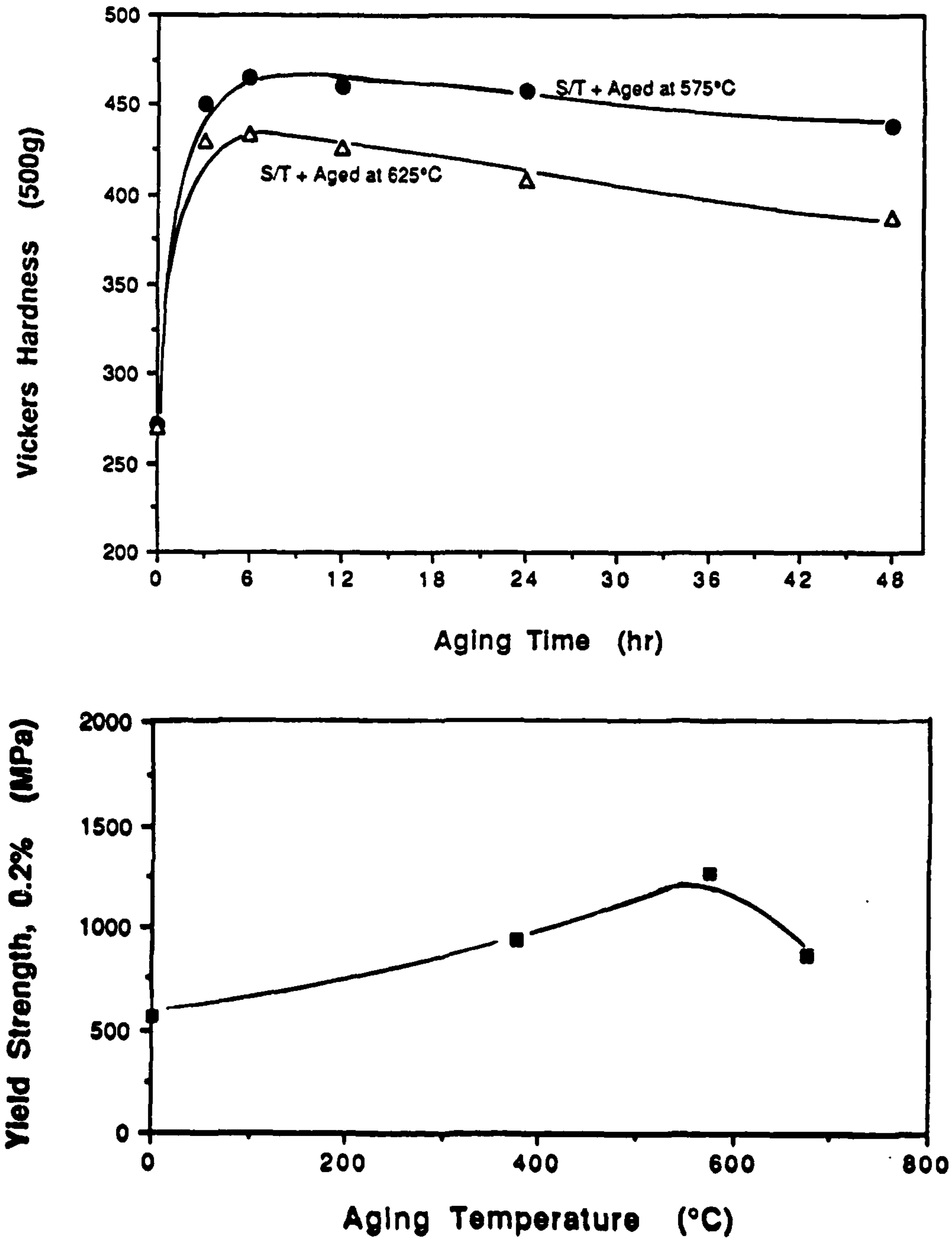

Figure 2. The age-hardening response of the Ti-23Nb-11Al after quenching from 1100°C.

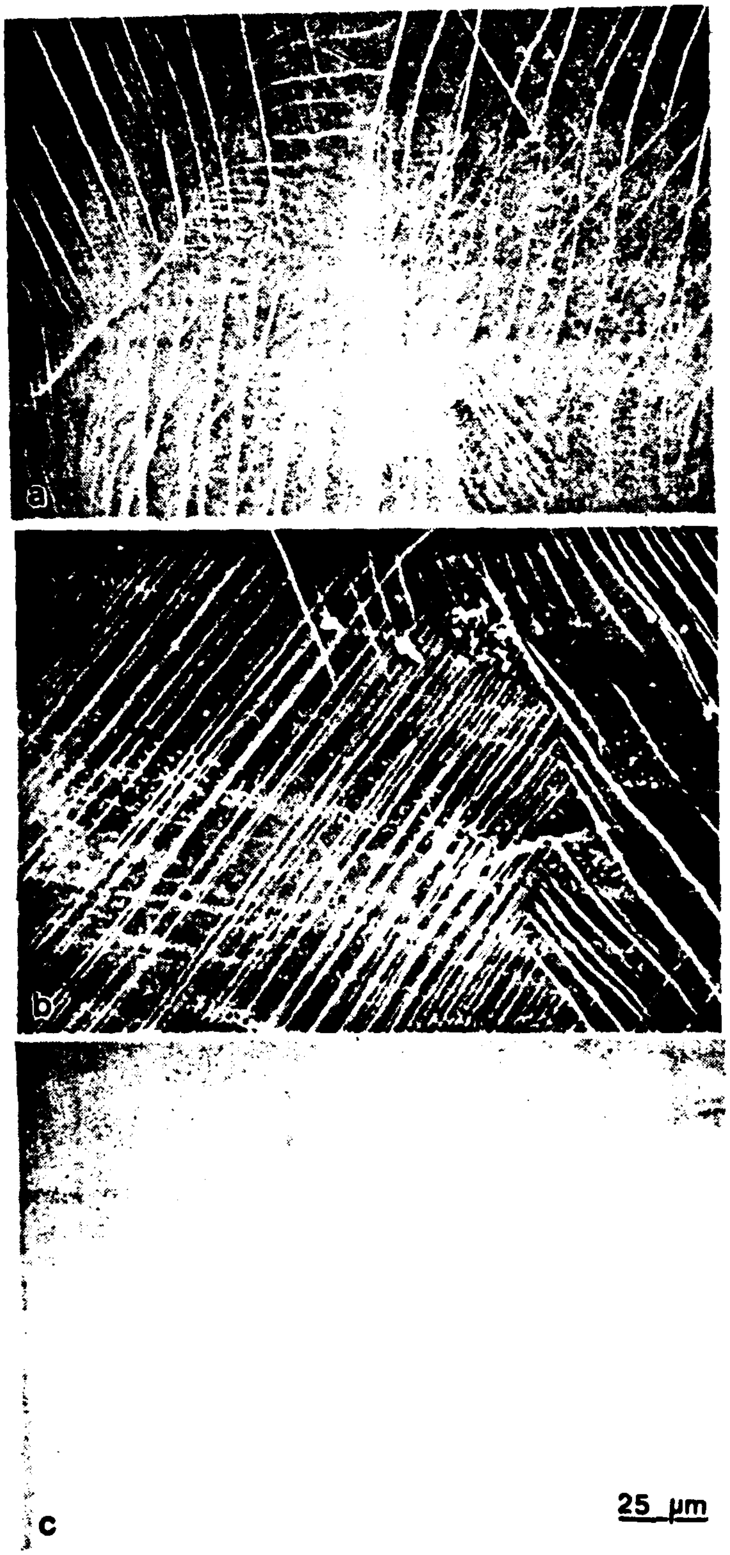

Figure 3. Optical micrographs showing slip lines on the surface of compression specimens after 0.03 strain and which have been solution-treated and aged for 6 hrs at (a) 375°C, (b) 575°C, (c) 675°C.

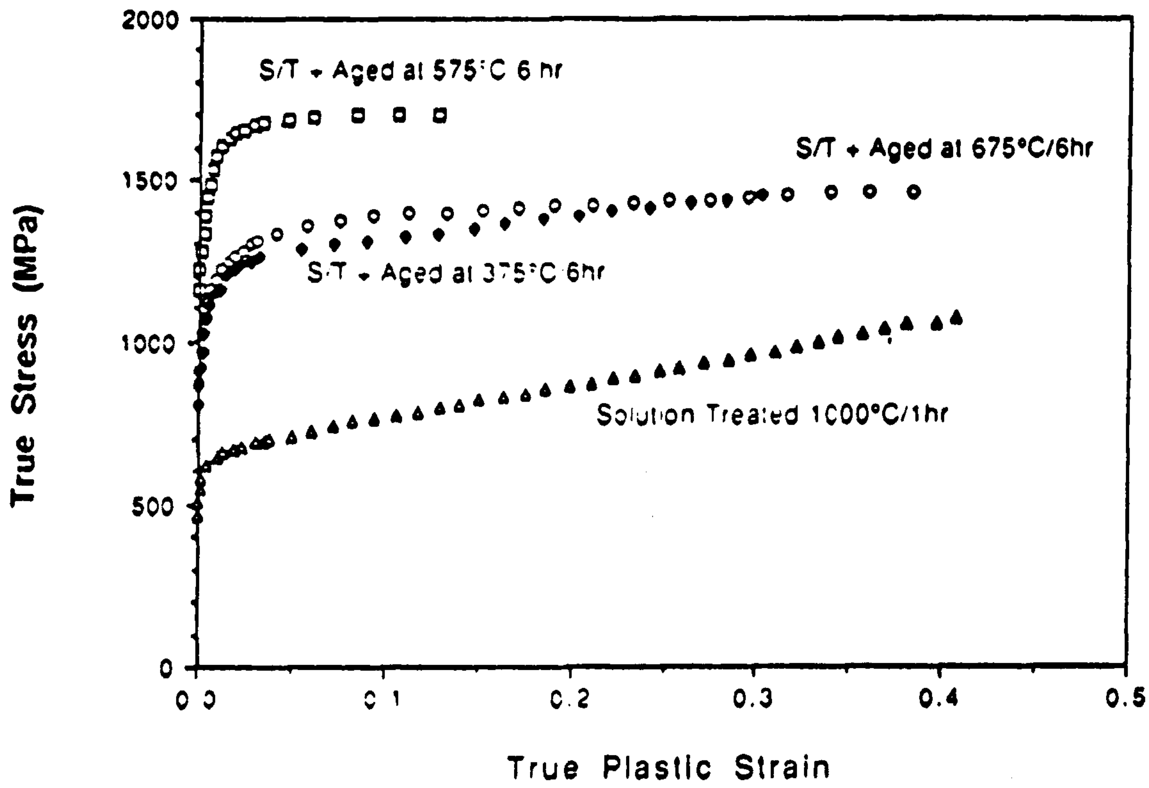

Figure 4. True stress - true strain behavior of Ti-23Nb-11Al specimens deformed in compression at 25°C.

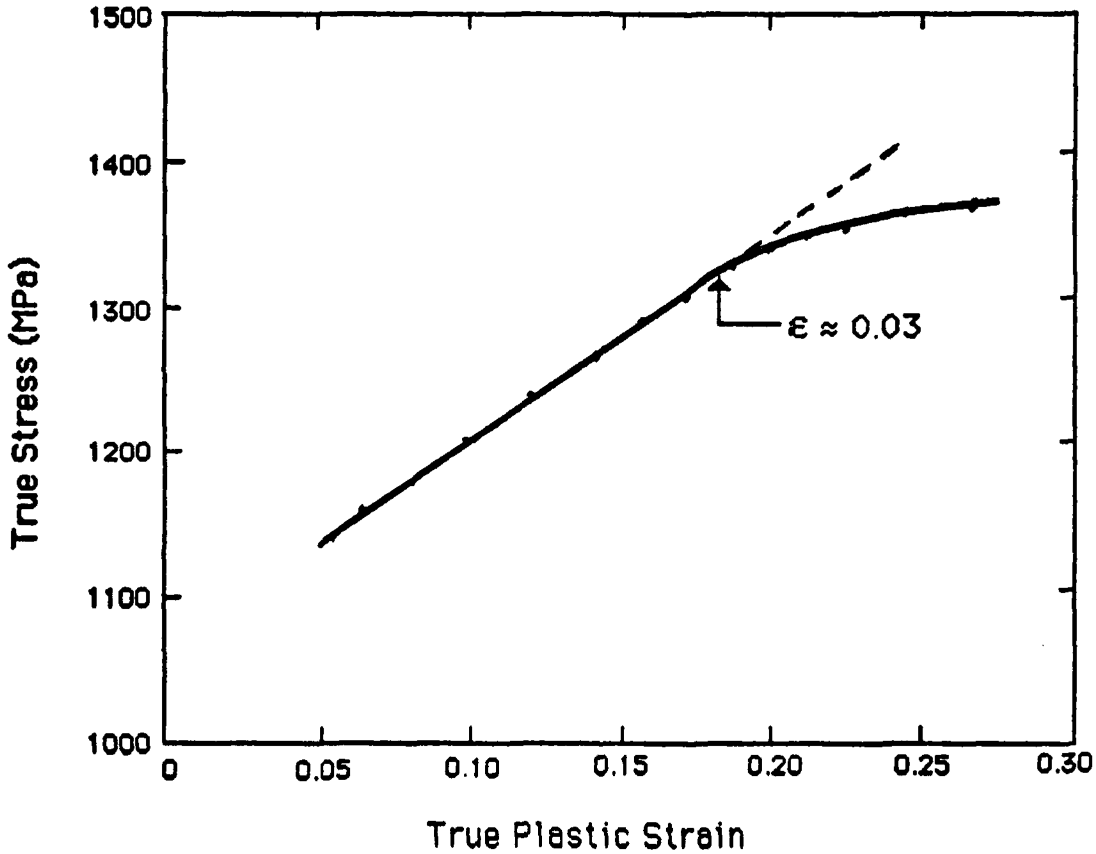

Figure 5. The dependence of the flow stress σ on the square root of plastic strain ε, as predicted by Eq. 1.